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**MATERIALS PROPERTIES - INFORMATION
FROM NON-DESTRUCTIVE AND DESTRUCTIVE
EXPERIMENTS VIA SIMULATION**

GEORGE N. FRANTZISKONIS

**DEPT OF CIVIL ENGR & ENGR MECHANICS
UNIVERSITY OF ARIZONA
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DECEMBER 1995

FINAL REPORT FOR 7/27/94--12/27/95

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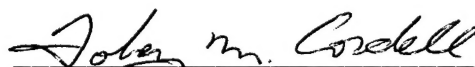
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13. ABSTRACT (Maximum 200 words) This work addresses procedures to obtain robust information on the mechanical properties of materials via simulation of relevant nondestructive and destructive experiments. Under this grant the general problem of information on properties of the fiber-matrix interface in composite materials is examined. A novel procedure for obtaining quantitative information on mechanical, including failure, properties of the interface is utilized, i.e., by simulating actual experiments in detail, including fiber properties of the interface is utilized, i.e., by simulating actual experiments in detail, including fiber breakage, matrix, yield and/or cracking, and interface failure. In a recent relevant study, the procedure was implemented for simulating commonly performed experiments, i.e., the fragmentation test for metal matrix composites (MMCs), the pushout and pullout tests for MMCs as well as ceramic matrix composites (CMCs). Herein the so-called transverse test is examined in detail with emphasis given on residual stresses. One major finding is that the (transverse) properties of the interface depend strongly on the residual stresses present, which, in turn, depend on the geometry of the structure/specimen examined and on the processing temperature.				
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OBJECTIVES OF THE RESEARCH EFFORT

The objectives are:

- To examine/establish a novel method for understanding the mechanisms involved during non-destructive and destructive testing of (composite) materials better.
- To utilize the technique as a tool for close cooperation between diverse materials evaluation groups, i.e. to establish concurrent analytical, computational, non-destructive and destructive research on composite materials.

Both objectives have been scrutinized, and the results have been very encouraging. Detailed examination of the "transverse" test (details are given below) through computer simulations has shown clearly the importance and relevance of the first objective. Since the simulation process makes possible the "visualization" of the mechanisms undergoing on the interface subjected to load, all groups listed under the second objective have interest in utilizing the technique. For example, as detailed in the sequence, it is possible, using the technique, to investigate and visualize the influence of residual stresses. Since this influence depends on the processing temperature as well as on the geometry of the specimen/structure examined, groups/individuals working on the processing aspect of the composite materials examined have interest in utilizing the simulation technique.

STATUS OF THE RESEARCH EFFORT

Soon after this grant was awarded, it was decided to concentrate on the so-called transverse test which aims at the mechanical properties of the fiber-matrix interface in, mainly, MMCs. The reason for this decision was the availability of unique and very relevant nondestructive and destructive tests performed at the Materials Laboratory, Wright-Patterson Air Force Base [1]. Based on [2], the test configuration (the transverse test described in detail in the sequence) is discretized into a lattice which delineates the matrix, the fiber, and the interface. The simulations provide further understanding of the mechanisms involved during the relevant testing. In addition, through back-analysis, quantitative values of the, homogenized, interface properties can be obtained.

In addition to the response of the interface to mechanical loading, the influence of residual stresses have been examined in detail. In an earlier study [3] the effect of specimen geometry on the residual stress distribution was examined, mainly near free surfaces. The simulations have shown that the specimen geometry as well as the processing temperature are key factors in determining the interface response. Especially for fracture initiation, the residual stresses are more often than not the most crucial governing factor. As an important consequence, definition of interface properties independent of residual stresses is of dubious merit. The present study suggests strongly that the interface cannot (should not) be examined in the context of local continuum theories, but rather as an entity closely related to the specimen/structure geometry it is embedded in as well as to the processing temperature. The simulation results agree closely with the analytical solution [3] where the effect of residual stress near boundaries is addressed. Further, the technique can be applied to any problem independent of the existence of an analytical solution. Detailed results are presented in the relevant publications.

BACKGROUND

There is a recent effort to investigate the transverse properties of titanium matrix composites (TMC) reinforced with continuous silicon carbide (SiC) fibers, and this involves understanding of the dependence of the fiber-matrix debonding on the residual stresses, on the fiber-matrix bond strength, and on the matrix properties, under transverse loading conditions of the composite. Novel methods for testing, nondestructively as well as destructively, the damage evolution mechanisms of a transversely loaded sample have been developed [1]. In these experiments samples with a single fiber were used, Fig. 1.

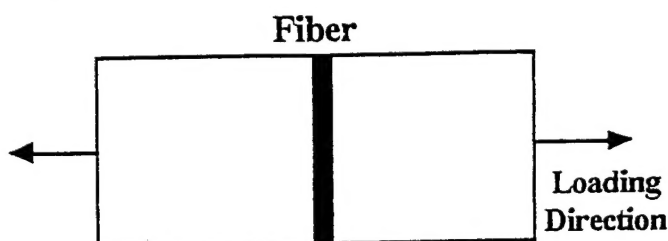


FIG 1: Schematic of Transverse Test Configuration Showing Orientation of Fiber in a Sample and the Direction of Loading

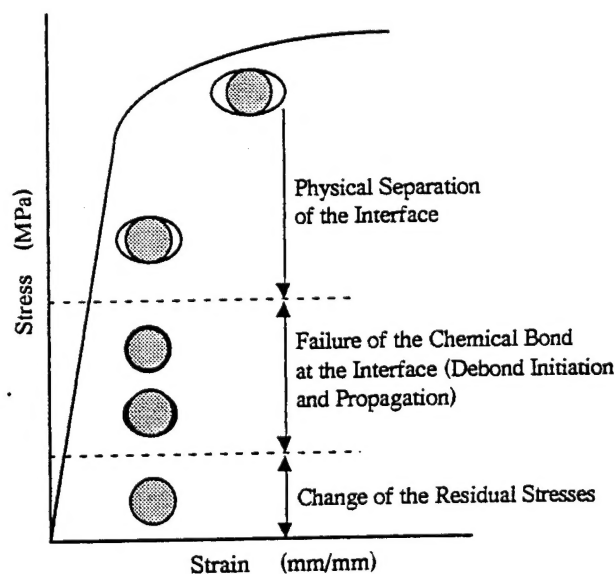


FIG 2: Evolution of Damage During Transverse Loading of a Composite Specimen.

The evolution of interface damage during the transverse test is illustrated schematically in figure 2, [1] and references cited therein. During the initial stages of loading, the radial compressive residual stress (not near boundaries) at the interface decreases while the interface remains intact. As the external loading increases further progressive failure of the chemical bond occurs, starting from the two sides in the direction of loading and progressing to the entire interfacial debond.

Figure 3 shows [1] the stress strain response obtained from a typical transverse test specimen, and figure 4 [1] shows the "Reflectivity Amplitude" at various stages of loading, as obtained by relevant ultrasonic testing. The higher the reflectivity, the more debonded the interface. Thus, figures

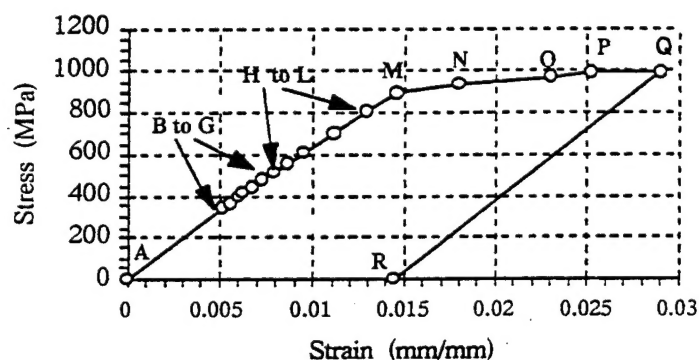


FIG 3: Stress-Strain Response of Specimen Subjected to Transverse Loading

3 and 4 show clearly that interface failure does not occur simultaneously nor homogeneously. Rather, the process of failure is highly heterogeneous along the length of the fiber, and no specific "heterogeneity pattern" can be identified.

The numerical simulations of the transverse tests, cf. Appendices agree both qualitatively and quantitatively with the above experimental observations, described briefly herein. Some limitations of the simulation approach, at the present stage of development, are discussed in the next section as well as in the appendices.

LIMITATIONS AND FUTURE WORK

As is shown in both appendices, A and B, the numerical simulations are performed using a so-called lattice analysis. Undoubtedly, a three dimensional analysis is preferable. This is especially true for simulation of the transverse test, since, due to boundary effects in the vicinity of the fiber edges, a

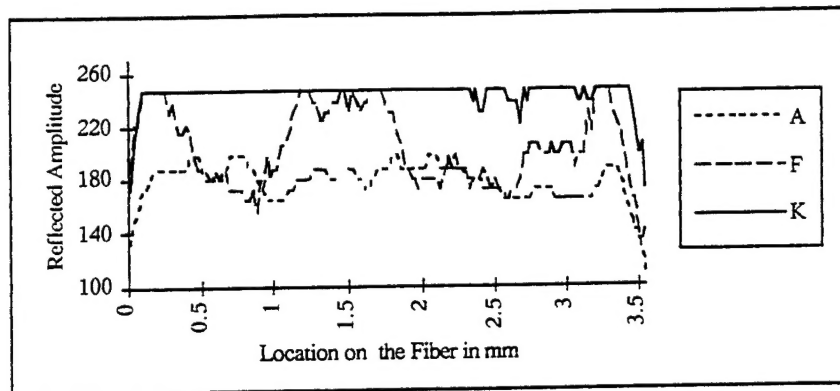


FIG 4: Spatial Along the Fiber Variation of Reflectivity Amplitude at Various Stages of Loading Indicated in Figure 3

representative plane stress or plane strain analysis of the problem cannot be pursued/identified. However, a fully three-dimensional lattice analysis is not trivial. This is due to the "unacceptable" computer time required for 3-D lattice analyses. The problem can be partially resolved by using either ultra fast single processor computers, i.e. large scale super computers, or, preferably, parallel (massively) computing.

Utilizing advanced computational facilities and techniques was out of the scope of this project. However, since even the 2-D analysis pursued shows promising results, it seems only natural to extend the present work to fully 3-D analysis. This would alleviate some of the problems encountered, cf. Appendices.

PUBLICATIONS

G. Frantziskonis, Theodore E. Matikas, Prasanna Karpur, S. Krishnamurthy, and Leon Shaw, Lattice Analysis to Assess Fiber-matrix Interface Behavior under Various Experimental Configurations, In: Computational Mechanics '95, S.N. Atluri, G. Yagawa, T.A. Cruse, Eds., Vol 2, pp 2563-2568.

The above publication is listed in APPENDIX A

G. Frantziskonis, T.E. Matikas, P. Karpur, The Effects of Residual Stresses on the Interface Response in MMCs, Simulation and Experiments, paper under preparation, to be submitted to *Composite Structures* or other relevant Journal.

The major results for the above publication appear in APPENDIX B

INTERACTIONS AND ADVISORY FUNCTIONS

The first of the two papers listed above was presented by George Frantziskonis at: Computational Mechanics '95, International Conference on Computational Engineering Science, July 30, 1995, Hawaii.

The project was completed under continuous interaction with the Materials Laboratory, Wright Patterson Air Force Base, Ohio. The PI, George Frantziskonis, interacted continuously with scientists there, cf. following names. He visited WPAFB in July 1995 to continue the cooperation and interacted with:

Dr. Thomas Moran, WL/MLLP
Dr. Theodore Matikas, UDRI, on site at WL/MLLP
Dr. Prasanna Karpur, UDRI, on site at WL/MLLP

One graduate Research Assistant, Mr. Gang Hong has been supported from this project. Part of the results presented herein are to appear in his Dissertation, scheduled for final examination during Fall 1996.

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- [1] L.L. Shaw, T.E. Matikas, P. Karpur, S. Hu, and D.B. Miracle, A Novel Method of In-situ Assessment of Fracture and Deformation of the Fiber/matrix Interface in Metal-matrix Composites, Proc. of the 1st Intl. Conference on Composites Engineering (ICCE/1), pp 479-480, New Orleans, 1994.
- [2] G.N. Frantziskonis, P. Karpur, T.E. Matikas, S. Krishnamurthy, and P.D. Jero, Fiber-matrix Interface - Information from Experiments via Simulation, Composite Structures, Vol. 29 (1994), pp. 231-247.
- [3] S. Hu, P. Karpur, T.E. Matikas, L. Shaw, and N. Pagano, Free Edge Effect on Residual Stresses and Fracture of Composite Fiber/Matrix Interface, Proc. of the 1st Intl. Conference on Composites Engineering (ICCE/1), pp 221-222, New Orleans, 1994.

APPENDIX A

Computational Mechanics ' 95

Theory and Applications

Proceedings of the International Conference on
Computational Engineering Science
July 30 - August 3, 1995, Hawaii, USA

Volume 2

Editors: S.N. Atluri, G. Yagawa, T.A. Cruse

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Lattice Analysis to Assess Fiber-matrix Interface Behavior under Various Experimental Configurations

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1. INTRODUCTION

This work utilizes a novel procedure for obtaining quantitative information on the mechanical properties of the fiber-matrix interface in composite materials. The method simulates actual experiments in detail, including fiber breakage, matrix yield and/or cracking, and interface failure. In a recent study, the procedure was implemented for the following commonly performed experiments: (a) the fragmentation test for metal matrix composites (MMCs); (b) the pushout and pullout tests for MMCs as well as ceramic matrix composites. In the simulations, the test configuration is discretized into a lattice which delineates the matrix, the fiber, and the interface. Details can be found in [1]. The simulations provide further understanding of the mechanisms involved during the relevant testing. In addition, through back-analysis, quantitative values of the homogenized, interface properties can be obtained. In this paper, we first describe simulations of the pushout/pullout and fragmentation tests for a titanium matrix, silicon carbide (SiC) fiber composite. Relevant interface properties are evaluated by simulating the former test. Using these values, we study the response of both test configurations, and then compare the numerical results with actual experimental data. Further, we combine recent experimental results with relevant simulations for the so-called transverse test of the same material composition.

2. FRAGMENTATION TESTS

The tensile load is applied in the direction of the fiber. Load is transferred from the matrix to the fiber, which at some point of loading breaks. Further loading results in the fiber breaking into successively smaller fragments until the fragments become too short to enable further increase in the fiber stress level. Since it is the fiber-matrix interface that actually delivers the load from the matrix to the fiber, the fragmentation "pattern" depends strongly on the properties of the interface. It is not, however, the interface that solely governs the fragmentation pattern. Relevant simulations [1] suggest, in addition to the importance of the interface properties discussed in the next paragraph, that: (a) the fiber fragmentation pattern depends strongly on the volume of the matrix present, (b) the matrix hardening modulus E_{mh} influences the fragmentation pattern, (c) the overall load-displacement response of the specimen is practically insensitive to the interface properties, (d) it is practically impossible to achieve average fragmentation lengths of the order of or lower than the fiber diameter.

In the numerical analysis, the transition region between fiber and matrix is modeled as an "interphase." Two values are necessary to describe its elastic brittle-

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response - the interphase stiffness coefficient S (modulus over thickness) and the failure stress coefficient F (failure stress over thickness). From back-analysis of the fragmentation test, the values for Ti-6Al-4V matrix, SiC (SCS-6) fiber composite were identified as: $S=860$ MPa/fm, $F=11$ MPa/fm. It is noted that the fragmentation pattern depends greatly on F , and not as much on S . For example, the fragmentation pattern obtained with the above values [1], is insensitive (within 100% changes) to the value of S , but depends strongly on F .

2. PUSHOUT AND PULLOUT TESTS

Such tests are commonly performed on metal as well as ceramic matrix composites. The specimens contain a single fiber with its ends exposed at both ends. In the pushout test an indenter loads the fiber in compression until it slides out of the matrix. The pullout test is similar, with the difference that tensile load is applied on the fiber. For metal matrix composites our simulations showed the following: (a) small, stable cracks form at the interface near the external load application side, for both the pushout and pullout tests, (b) after the arrest of the cracks mentioned above, tension and shear dominated interface cracks initiate at the side opposite to the external load application, (c) after the arrest of the cracks that initiated at the bottom, further external load is required for the whole interface to fail, (d) the size of the "hole" at the bottom support has significant effect on the load-deformation response, and (e) the effects of residual stresses seem to be important. Details can be found in the literature [1].

3. TRANSVERSE TESTS

The transverse test which is used to study the interface fracture behavior is usually performed using single fiber model composites. Single fiber specimens are well suited for interface evaluation because, (a) the interface chemical bonding, which depends on the chemical reaction between the matrix and the fiber materials during processing, remains the same in a single fiber sample as well as in a multi-ply composite panel, and (b) the residual stresses at the interface are relatively easy to calculate in the single fiber sample, thereby making it feasible to account for the residual stresses in the modeling of the test to derive more accurate conclusions about the stress at fracture of the fiber-matrix interface. In the following we first describe some experimental results from recent transverse tests [2] and then the results from the lattice modeling studies.

Monofilament composite samples for these experiments were processed by the foil-fiber-foil technique wherein two Ti-6Al-4V sheets with a single SiC fiber (SCS-6 or SCS-0) between them were hot pressed at a condition of 960 °C at 17 MPa for 1.5 hours. The interfacial microstructure obtained with this processing condition clearly indicates that some chemical reaction between the graded carbon coating and the matrix has taken place during the consolidation process. Furthermore, this reaction is nonuniform. After processing of the single fiber composite samples, the samples are cut into dog-bone-shaped specimens with the fiber axis perpendicular to the loading axis of the samples as shown in Figure 1a.

3.1 Experimental Approach

Transverse tensile tests were carried out using a micro-straining stage built in the WL/Materials Directorate [3]. The loading was applied stepwise so that the ultrasonic scanning could be done in-situ under the loaded condition at different stress levels. An in-situ ultrasonic nondestructive technique was used for this purpose [3].

The damage of the fiber-matrix interface during the transverse test was evaluated using the shear-wave back reflectivity technique (SBR) [4, 5]. The loading of the

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samples was done in incremental steps. At each step of loading, the fiber-matrix interface was imaged (while holding the sample under that load) in a pulse-echo mode using a focused ultrasonic beam which was incident on the surface of the samples at an angle of 24° , an angle between the first and the second critical angles of the matrix material. Hence, mode-converted vertically polarized shear waves were incident on the fiber-matrix interface. The shear waves were back-reflected to the transducer and the reflection coefficient from the interface was evaluated to characterize the interface fracture. The model presented in literature [4, 5] was used for the modeling of the reflection coefficient and the interface fracture was modeled based on theoretical predictions and experimental data. The reflected amplitude from a hole, which represents a complete debond (see previous section), was used to calibrate the maximum reflected signal from the fiber to correspond to the fracture of the interface.

Figure 1b shows the stress-strain diagram for the tensile test of a Ti-6Al-4V/SCS-6 single fiber composite sample under transverse loading, and the corresponding ultrasonic shear wave images at various stress levels of the sample labeled 'A' through 'K' in the Figure. The image labeled 'A' in Figure 1b corresponds to the fiber-matrix interface before the commencement of the mechanical loading of the sample. Reflectivity image 'B' indicates the first few points of interface that fracture at about 350 MPa (shown with the maximum amplitude calibrated to red in the color scale). These points of interface fracture are located near the two ends of the fiber and in several places at the center of the fiber. The reflectivity images 'C' through 'J' show the progression of interfacial damage as the load increases. The image 'K' shows that the entire interface has been fractured at about 700 MPa. An important conclusion drawn from using in-situ SBR imaging of the transverse test contradicts a commonly accepted assumption that the entire interface is likely to fracture almost instantaneously once a sufficiently high stress level is reached because of the existence of a weak diffusion bond [6] (as contrasted to mechanical bonding). The work reported here suggests that the debonding progresses from a small number of isolated points at a low load, to the entire interface over a finite range of applied stress (almost 350 MPa as shown in Figure 1b). This range is dependent mainly on the redistribution of stress along the interface which occurs due to the propagation of the interfacial crack as well as on the homogeneity of the fiber-matrix interface. Finally, the applied load completely released and the image of the fiber significantly reduced as the open interface cracks closed.

3.2 Lattice Modeling of Transverse Tests

The test configurations mentioned above are three-dimensional. Thus a simulation should reflect the 3-D effects. In [1] the simulations were two-dimensional; the limitations from such a "simplification" are discussed extensively in [1]. For the fragmentation and pullout, pushout tests, since the load is applied parallel to the fiber, the limitations from a 2-D simulations may not be of great importance, although a 3-D simulation process would be preferable. The reason for performing 2-D simulations have been the following: (a) there are several issues to be resolved, i.e. thermally induced residual stresses, the implications of assuming a homogenized interphase, effects of temperature on interface properties, imperfections in the interface, fiber, and matrix, etc., before a fully 3-D simulation process is employed, (b) for 3-dimensional analysis, only problems with rather coarse discretization can be solved within reasonable computer time use. For the transverse test, the limitations of a 2-dimensional analysis may be important. However, certain issues can be examined by analyzing a configuration like the one shown in Figure 2a.

Figure 2b shows the crack pattern obtained from the lattice analysis, at the load level shown by an arrow in Figure 3. The following have been identified: (a) before the matrix yields, the interface starts to fail, (b) interface failure is not symmetric with respect to either the loading axis, or the axis perpendicular to the loading direction, (c) although interface failure is locally brittle, the final stage of interface failure is reached

progressively, (d) the matrix starts yielding only after "complete" failure of the interface, after that stage the matrix starts yielding and develops a "shear band" type failure as shown in Figure 4, (e) the external load increase needed to advance interface failure from initial to its final failure state is not negligible.

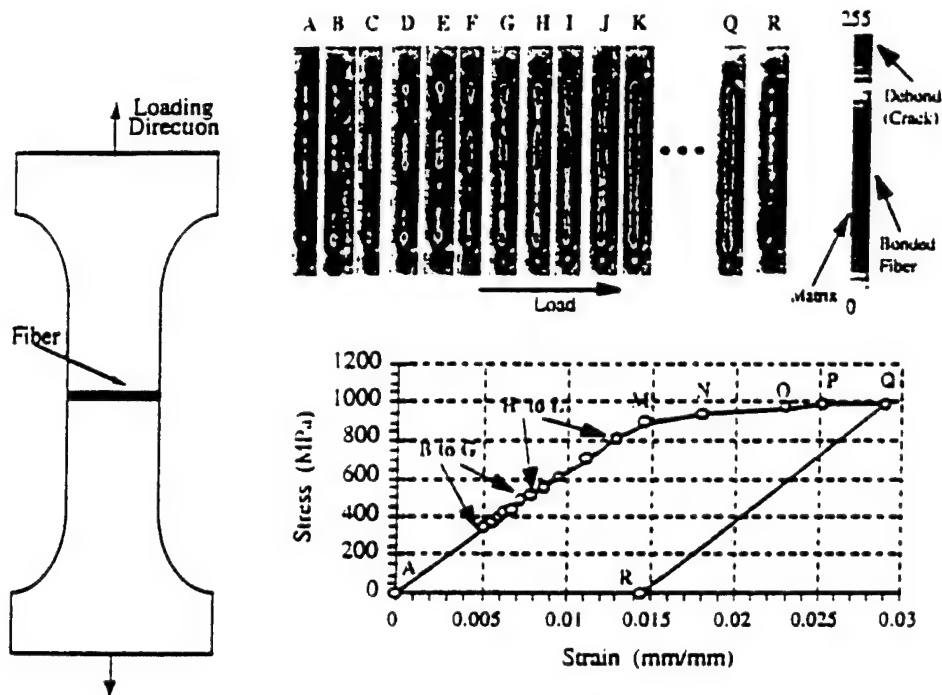


Figure 1 (a) Experimental Configuration Showing Transverse Orientation of the Fiber in a Sample and the Direction of Loading. (b) In-Situ Ultrasonic SBR Imaging of an Embedded SCS-6 Fiber in Ti-6Al-4V Matrix During Various Stages of Transverse Loading.

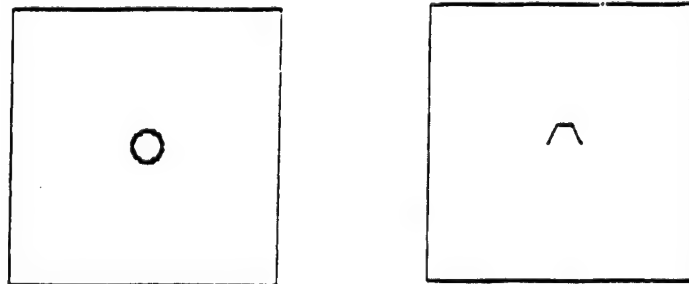


Figure 2: (a) Two-dimensional configuration of the transverse test. The bars in the lattice assigned interface properties are shown. The fiber is inside the "ring," and the rest of the configuration is matrix. (b) Interface crack pattern from two-dimensional simulation of the transverse test. The corresponding external load level is shown by an arrow in Fig. 3.

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In our simulations interface failure initiated at about 339 MPa external stress and progressed up to about 520 MPa. Thus it is difficult to define an external "stress" that corresponds to interface failure for this experimental configuration. Despite this difficulty in order to examine the response along the direction of the fiber, we have considered a "failure stress" of 420 MPa and with this value we performed simulations (2-D), as shown in Figure 5a, where now the x-axis is parallel to the fiber and the y-axis coincides with the external load axis. In these simulations, all interface elements are subjected to the same stress, thus we chose to initiate fracture at the right free-surface. We believe that residual stresses, which the present analysis does not consider, will be important for interface fracture along the fiber direction. Figure 5b shows the interface crack propagation at some point of external load application. Here interface fracture is "brittle" (the external stress increase from interface fracture initiation to its complete fracture is not appreciable).

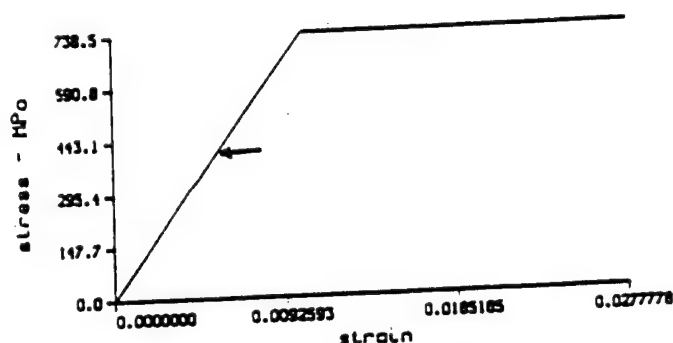


Figure 3: The external stress Vs external strain response as calculated from the two-dimensional simulation.



Figure 4: Simulation results. Failure (yielding) of the ductile matrix, occurring after interface failure. The corresponding external strain is 0.02.

Thus, for the transverse test, interface (overall) failure on the plane transverse to the fiber axis is non-brittle, while crack propagation in the direction of the fiber is rather brittle. This shows the strong interactions among the different directions. In other words, three-dimensional effects are important for the transverse test. Mean interface failure values can be obtained, however, by 2-D analysis using the configuration of Figure 2. It is once again noted that the simulations were performed using interface values as determined from the back analysis of the fragmentation test. Thus the results of the 2-D simulations can be compared with the mean (averaged over the fiber direction)

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experimental response. Averaging is necessary since the 3-D interactions may render interface failure highly non-uniform along the fiber direction. We are currently in the process of correlating experimental and simulation results, and on performing actual 3-D simulations.

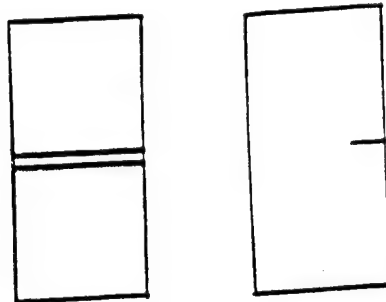


Figure 5: (a) Along the fiber axis two-dimensional configuration of the transverse test. The bars in the lattice assigned interface properties are shown. The fiber is inside the two parallel lines and the rest of the configuration is matrix. (b) Interface crack pattern soon after crack initiation.

ACKNOWLEDGMENTS

This work is partially supported by the Wright Laboratory, Wright-Patterson Air Force Base, Contract Nos. F33615-94-1-5210 (GF), F33615-94-C-5213 (PK, TEM), F33615-92-C-5663 (SK), and by the Hughes Aircraft company (GF).

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APPENDIX B

Two major improvements of the work presented in Appendix A have been investigated:

- Detailed study, using the lattice based simulation process, to determine the residual stress distribution, i.e. before any external mechanical load is applied on the specimen.
- Examination of the fiber-matrix interface properties including the residual stress distribution. In other words this second "improvement" calls for identification of the apparent interface properties which include the effect of residual stress.

Before presenting and discussing the results, the mechanical properties used for the titanium alloy matrix and for the SiC fiber are given in the following table. The processing temperature was considered to be 900°C.

	Density gr/cm ³	Young's Modulus GPa	Shear Modulus GPa	Yield (matrix) or Failure (fiber) MPa	Poisson's Ratio	Thermal Expansion Coeff 10 ⁻⁶ /°C
Ti-6Al-4V	4.5	119	45	830	0.32	9.0
SiC SES-6	3.2	411	179	3500	0.15	3.99

As mentioned above the analysis was restricted to two-dimensions. Thus, first the response of the specimen far from the fiber edges, i.e. for an infinite sized structure was analyzed by using a

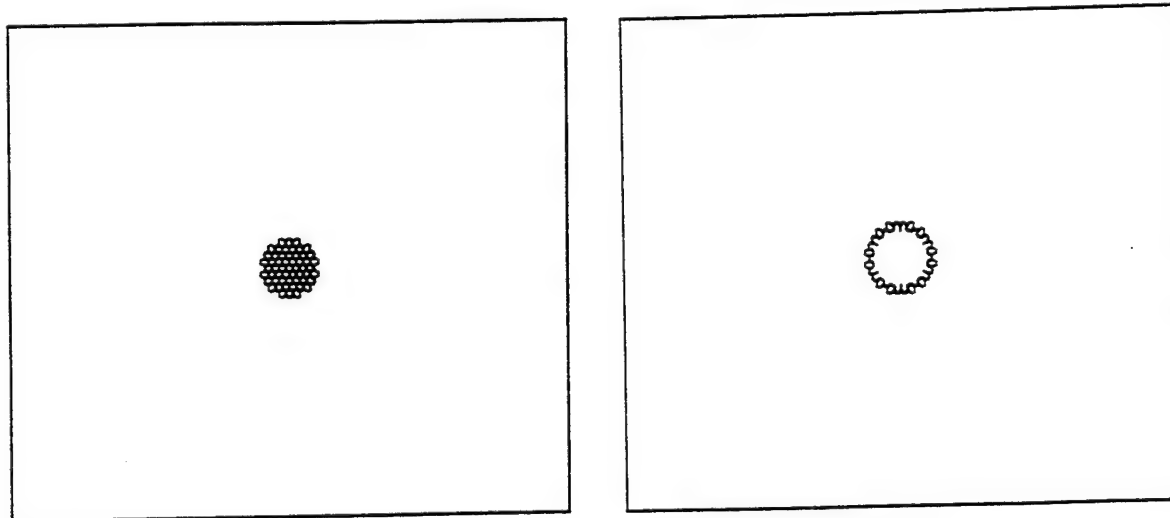


FIG B1: The First 2-D Configuration Analyzed. (a) 80x80 Lattice where the Bars Assigned Fiber Properties are Shown, (b) the Interface Bars are Shown.

configuration such as that of Fig. B1. The lattice sizes varied from 80x80 to 100x100. The dimensions of each specimen are 0.0015x0.0015m, the fiber diameter is 150 μ m, and the interface properties are considered (normalized/divided) per unit of its thickness, as explained in a lucid fashion in [2]. Thus, in the sequence all interface properties are given as values per thickness. In the configuration shown in Fig. B1, the interface thickness is 0.0015/80 m, that of one lattice spacing.

The matrix is considered elastic-plastic, the elastic part being linear and the hardening being also linear (elastic, linear hardening). In other words, the matrix is allowed to yield and develop high strains. The fiber is considered linearly elastic and brittle. After the failure stress is reached the load (i.e. the load of a fractured bar in the lattice analysis) is reduced to zero. The interface is also considered to be linearly elastic and brittle. Then, as discussed in detail in [2] the problem reduces in determining the elastic and failure properties of the interface. This is done by back analysis of the simulations described below and/or the values determined from back analysis of tests other than the transverse are used. For example for Ti-6Al-4V matrix, SCS-6 fiber composite the interface properties were determined in [2] by back analysis of the fragmentation, the pullout and the pushout tests.

Figure B2 shows six (6) views of the residual stresses before any external load is applied, i.e. due to the processing temperature difference. As can be seen, the residual stresses are quite high, i.e. as compared with the yield/failure stresses for the matrix/fiber. Note that this configuration, Fig. B1 corresponds to an infinitely long fiber, thus boundary effects are excluded. Thus the interface is mostly subjected to compressive stress. The elasticity properties used for the interface in obtaining the stress distribution shown in Fig. B2 are those obtained by back analysis of the fragmentation, pullout and pushout tests [2]. The so-called modulus coefficient (Young's modulus divided by interface thickness) is 857 MPa/ μ m. In [2] the interface failure strength coefficient was determined to be 11 MPa/ μ m. In the transverse test the thickness of the sample is 0.004m, thus the fact that the behavior near the edges is actually different (boundary effects) is important. Thus, failure is governed by the interplay of boundary and bulk (homogeneous) interface response. It was determined that the value of 11 MPa/ μ m for the interface failure strength coefficient needed to be adjusted for the (apparent) inclusion of boundary effects to 13.65 MPa/ μ m. Once again, problems of such "adjustments" would not appear if a fully three dimensional analysis was performed. Figures B2a and B2b show a 3-D plot of the residual stresses, σ_{xx} , for the configuration of Fig. B1. They show that a low tensile stress develops in the matrix, while a "high" compressive value develops at the interface and in the fiber. Figures, B3c-f show contour plots of the same stress distribution, σ_{xx} , included herein for the sake of clarity.

The residual stresses influence the process of failure of specimens significantly. For the configuration of Fig. B1 partial cracks of small extend develop in the interface during the early straining stages. However, these cracks did not propagate and cover only a small fraction of the total interface. Before further interface cracking occurs, a large part of the matrix yields. Figure B3 shows a typical interface fracture "pattern" at some stage of loading. From that point on, the interface failure is rather rapid and the total interface fails at an external strain of about 3%.

In order to study the distribution and the effects of residual stresses near the fiber ends, the configuration of Fig. B1 is inadequate. Again, a three dimensional analysis would be preferred. Here we "bypass" a fully 3-D analysis by studying the 2-D configuration like that of Fig. B4. Figure B5 shows different views of the residual stress σ_{yy} . The high tensile stresses near the edges are evident. This agrees with experimental observations which call for tensile microcracks being present at the

interface, near the edges of the specimen. Thus, the interface is already (before the external straining is applied) cracked near the specimen edges. This initial cracking, together with the residual stresses in the remaining of the interface influence the way the interface cracks under external straining of the sample. As shown in Fig. B6 subsequently to the initial cracking, interface cracks develop near the two sample edges as well as in the center. This agrees with the experimental observations described above.

Note that the configuration of Fig B4 is not representative of the "true" response, due to the limitations of a 2-D analysis. Despite this, note that the stress-strain response obtained from the lattice analysis agrees very well (quantitatively) with the experimental one, that of Fig. 3, page 4.

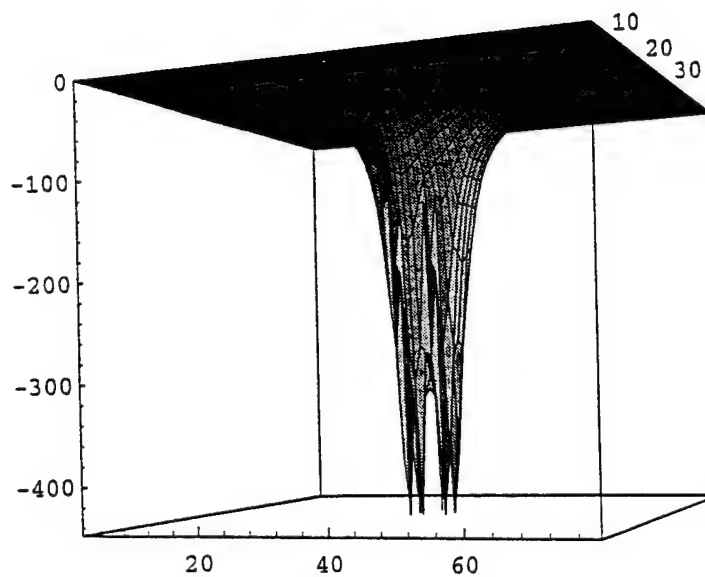
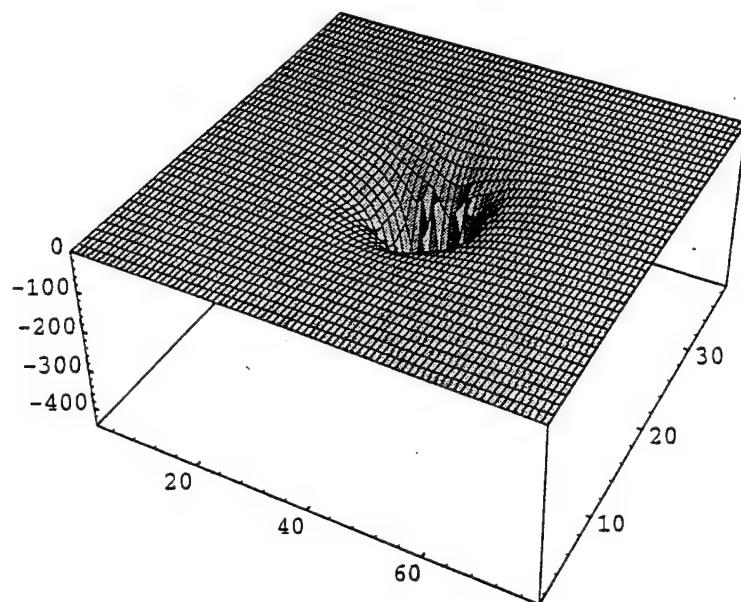


FIG B2a,b: Two Different (3-D) Views of the Residual Stress Distribution for the Specimen Described in Fig. B1. The Values of σ_r are Plotted in MPa.

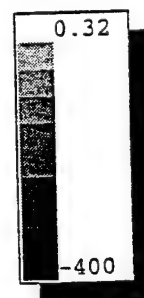
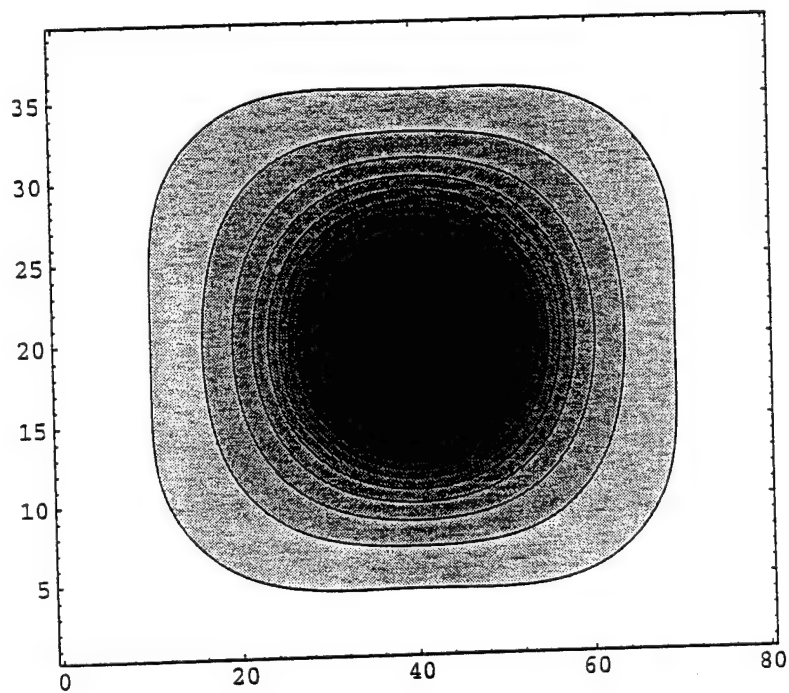
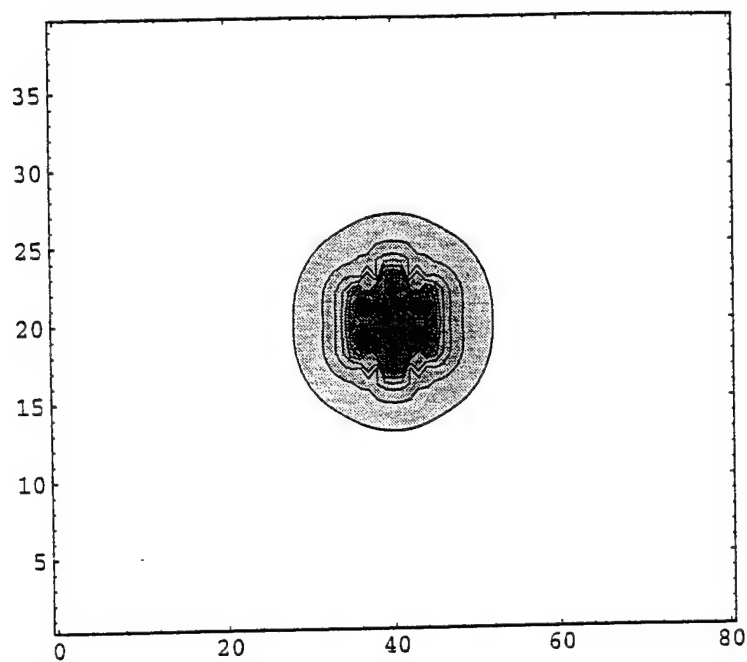


FIG B2c-d: Contour Plots of the Stress Distribution Shown in Fig B1a,b.

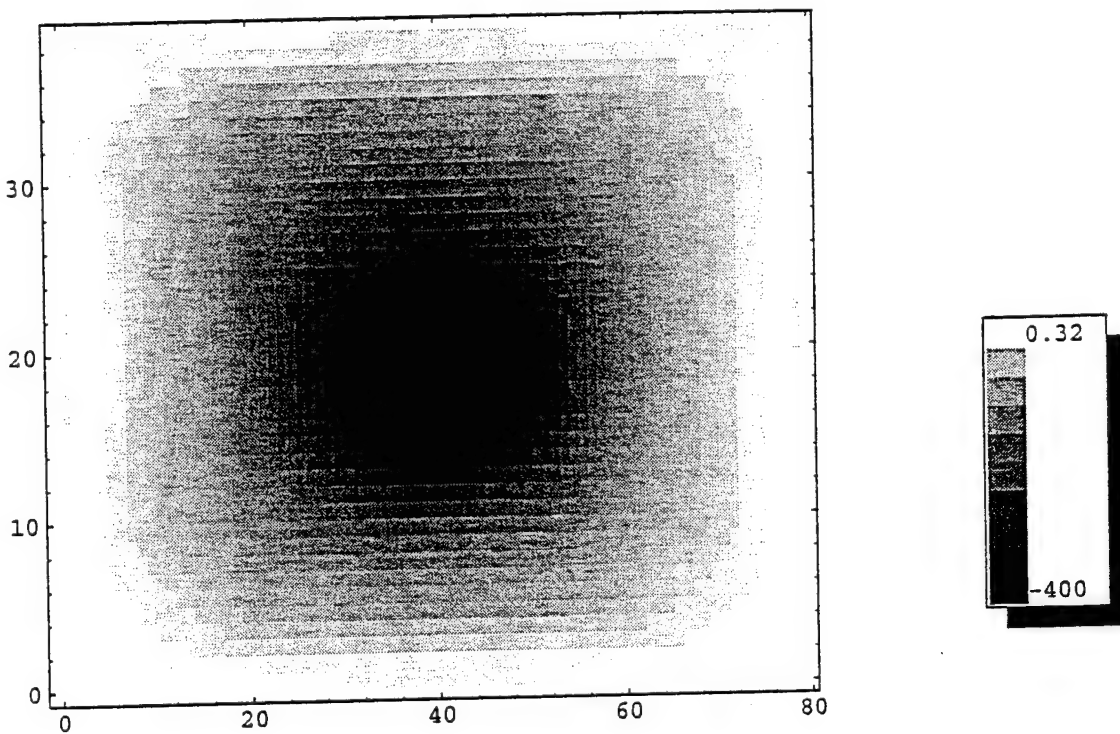
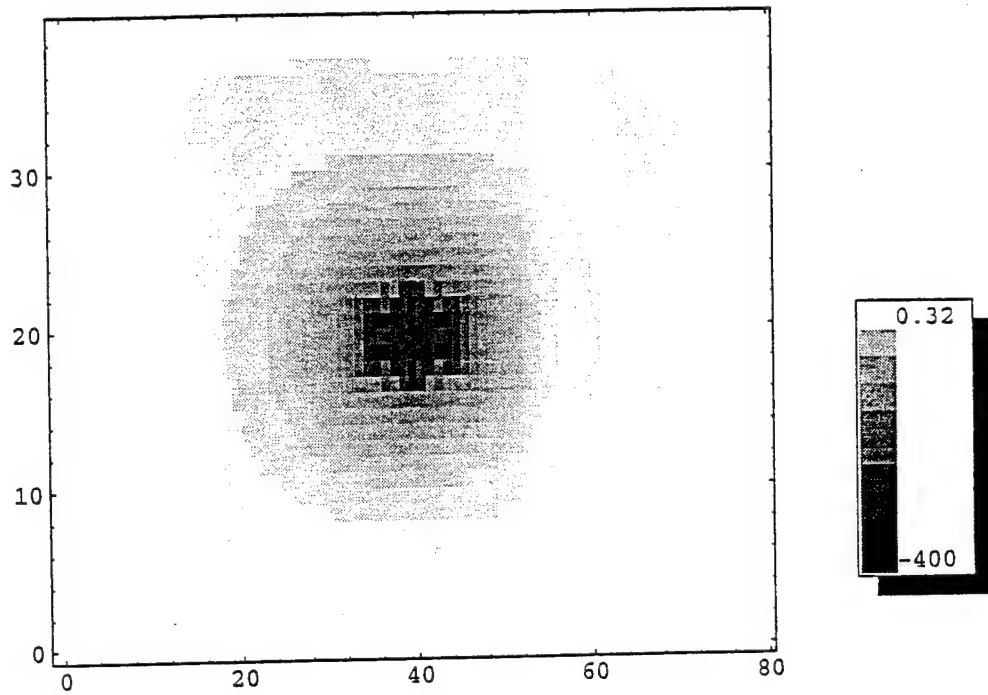


FIG B2e-f: Contour Plots of the Stress Distribution Shown in Fig B1a,b.

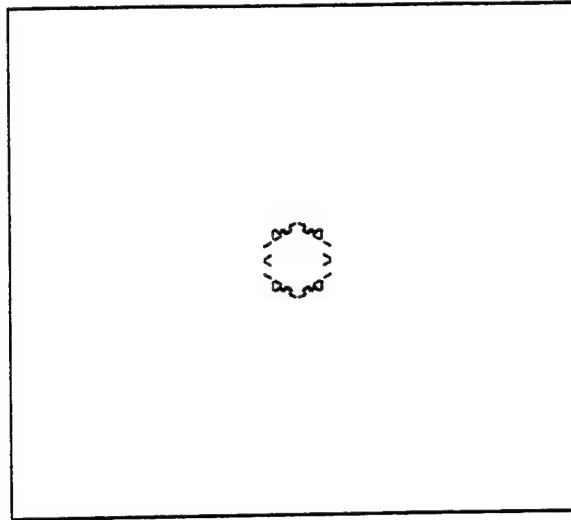


FIG B3: Interface Fracture Pattern at an Intermediate Load Step.

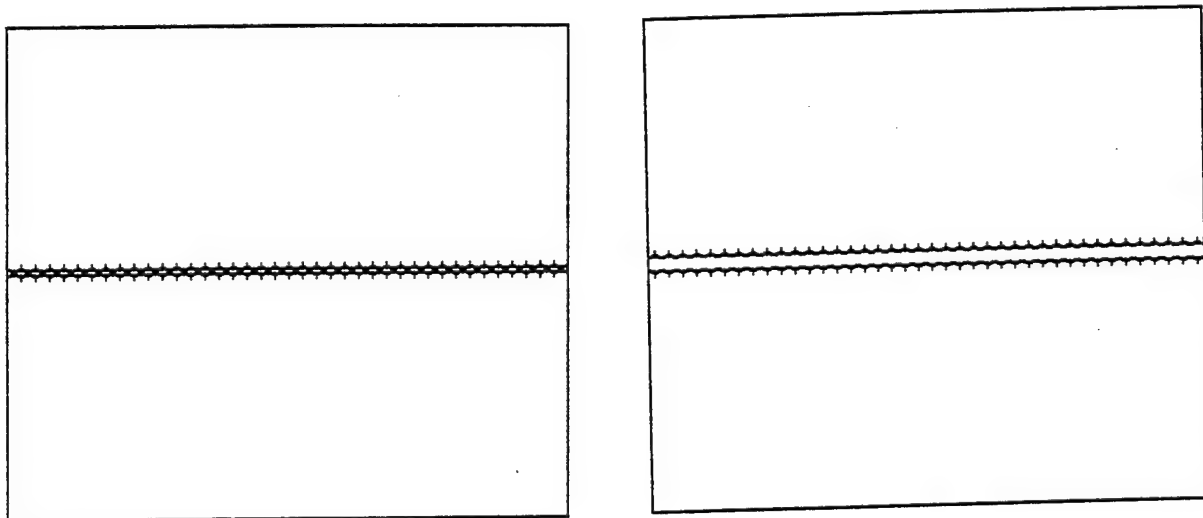


FIG B4: The second 2-D configuration Analyzed. (a) 40x100 Lattice where the Bars Assigned Fiber Properties are Shown, (b) the Interface Bars are Shown.

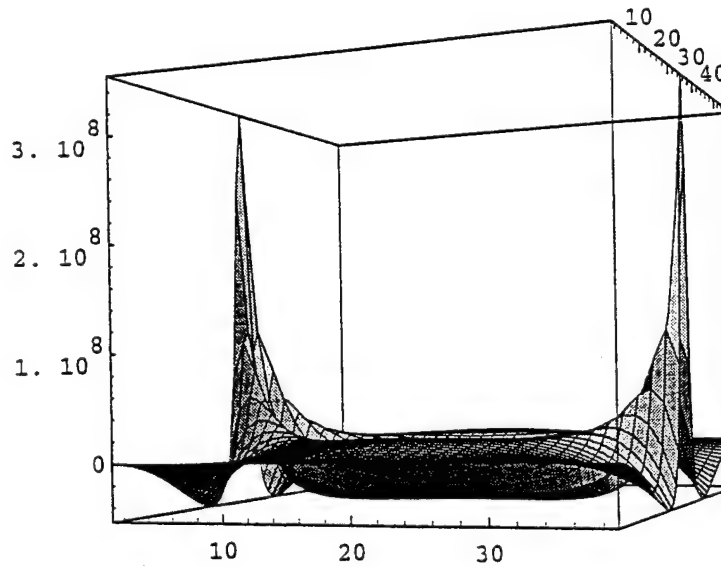
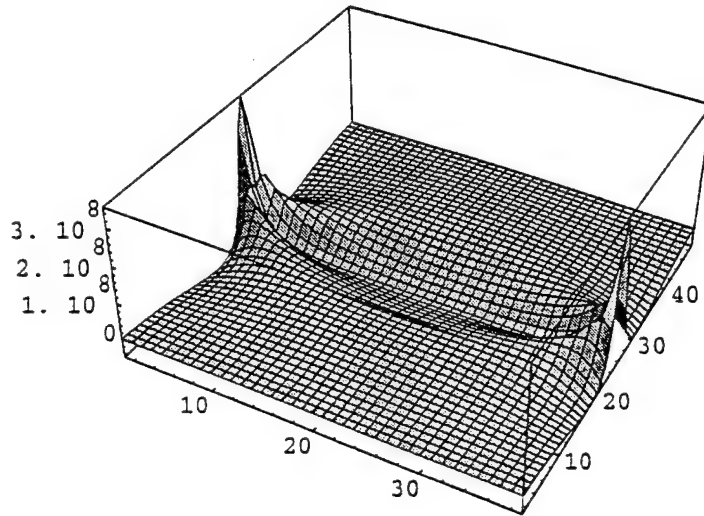


FIG B5a,b: Two Different (3-D) Views of the Residual Stress Distribution for the Specimen Described in Fig. B4. The Values of σ_{yy} are Plotted in Pa.

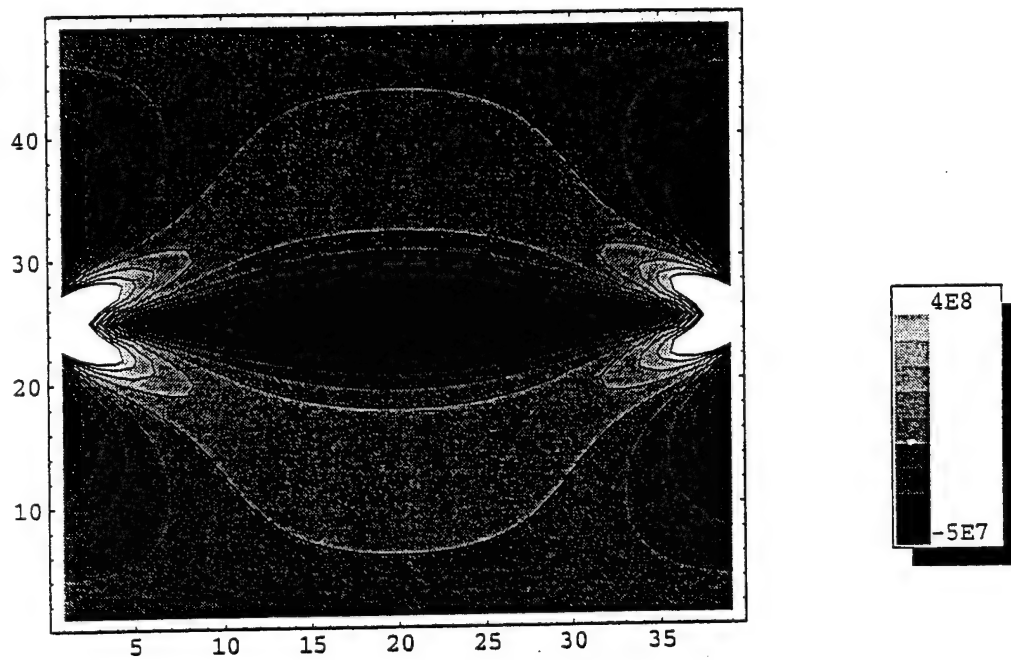
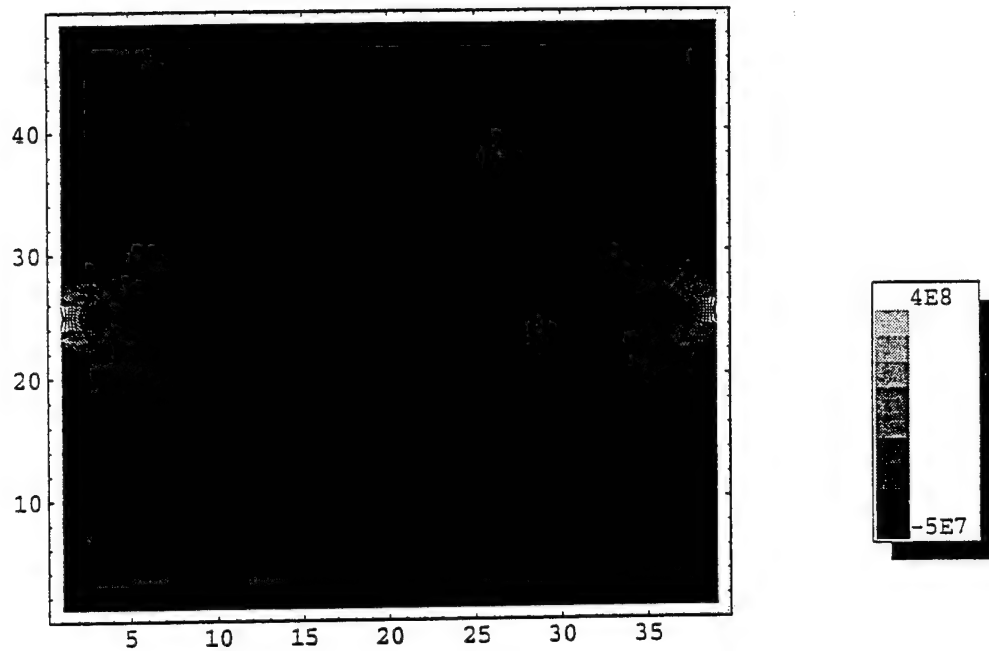


FIG B5c-d: Contour Plots of the Stress Distribution Shown in Fig B5a,b.

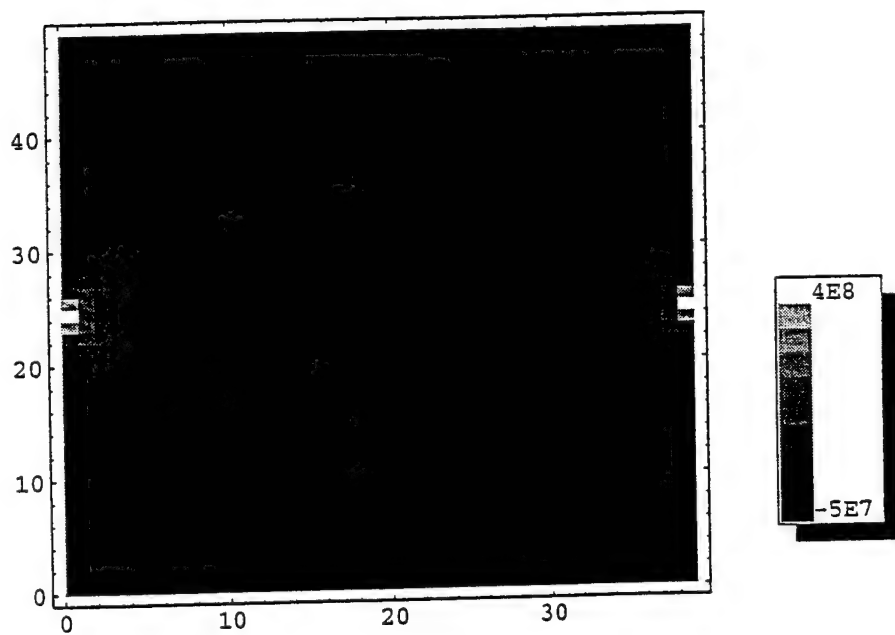
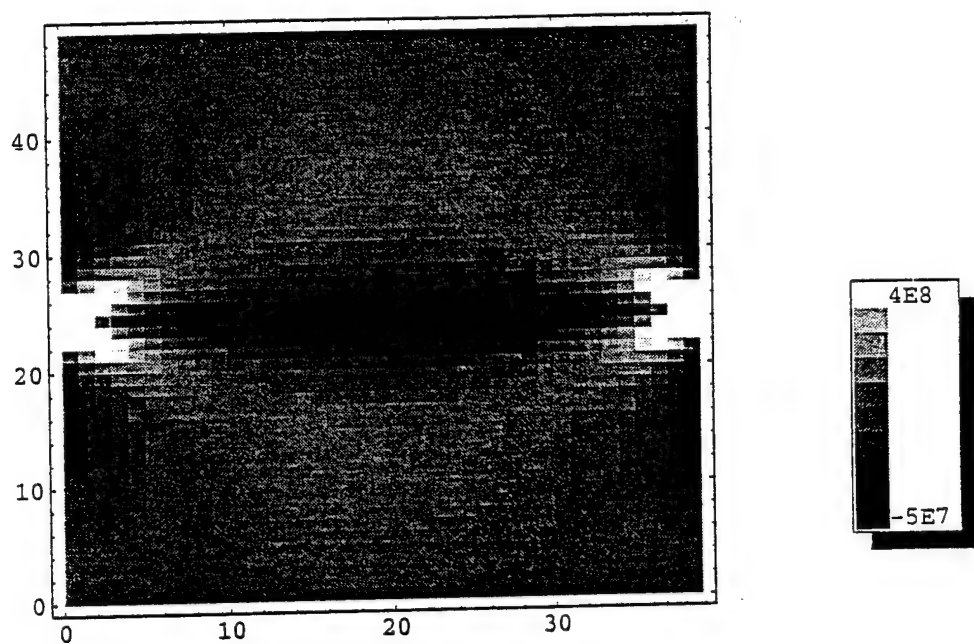


FIG B5e-f: Contour Plots of the Stress Distribution Shown in Fig B1a,b.

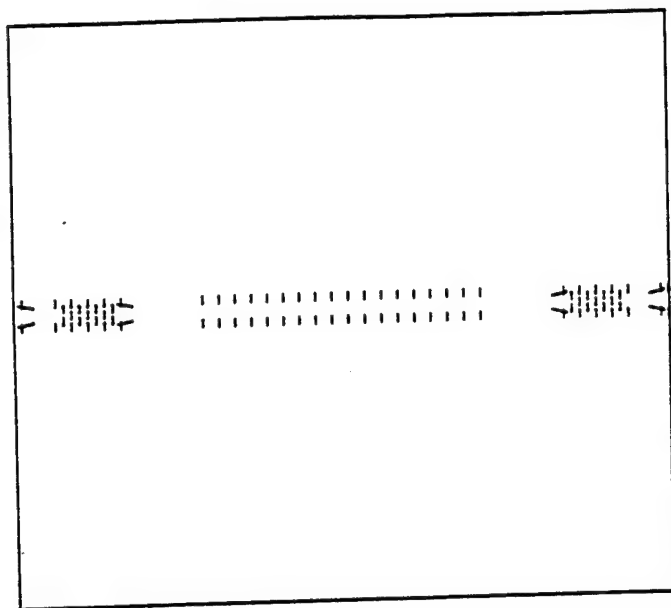


FIG B6: Interface Fracture Pattern at an Intermediate Load Step.

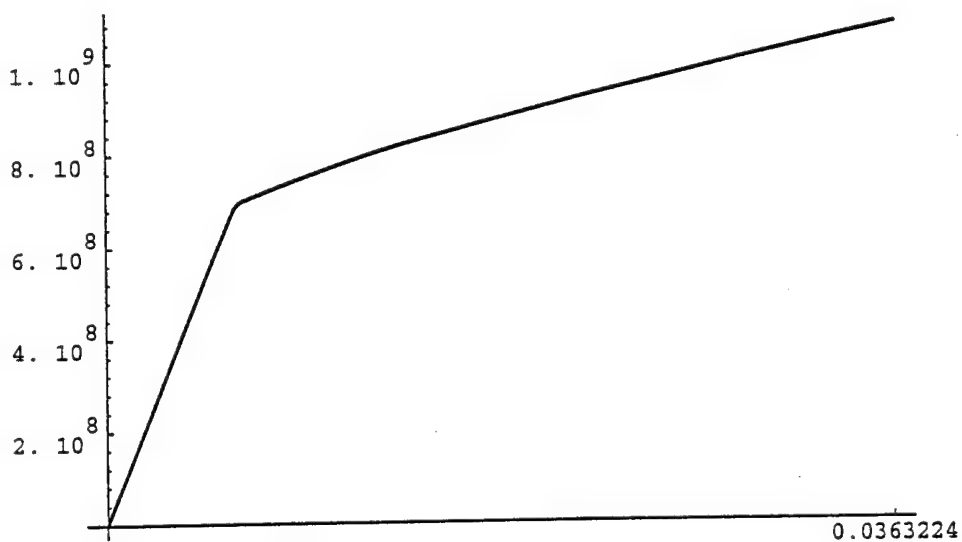


FIG B7: Stress (Pa) versus Strain, Obtained from the Configuration of Fig. B4.

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Lattice Analysis to Assess Fiber-matrix Interface Behavior under Various Experimental Configurations

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1. INTRODUCTION

This work utilizes a novel procedure for obtaining quantitative information on the mechanical properties of the fiber-matrix interface in composite materials. The method simulates actual experiments in detail, including fiber breakage, matrix yield and/or cracking, and interface failure. In a recent study, the procedure was implemented for the following commonly performed experiments: (a) the fragmentation test for metal matrix composites (MMCs); (b) the pushout and pullout tests for MMCs as well as ceramic matrix composites. In the simulations, the test configuration is discretized into a lattice which delineates the matrix, the fiber, and the interface. Details can be found in [1]. The simulations provide further understanding of the mechanisms involved during the relevant testing. In addition, through back-analysis, quantitative values of the homogenized, interface properties can be obtained. In this paper, we first describe simulations of the pushout/pullout and fragmentation tests for a titanium matrix, silicon carbide (SiC) fiber composite. Relevant interface properties are evaluated by simulating the former test. Using these values, we study the response of both test configurations, and then compare the numerical results with actual experimental data. Further, we combine recent experimental results with relevant simulations for the so-called transverse test of the same material composition.

2. FRAGMENTATION TESTS

The tensile load is applied in the direction of the fiber. Load is transferred from the matrix to the fiber, which at some point of loading breaks. Further loading results in the fiber breaking into successively smaller fragments until the fragments become too short to enable further increase in the fiber stress level. Since it is the fiber-matrix interface that actually delivers the load from the matrix to the fiber, the fragmentation "pattern" depends strongly on the properties of the interface. It is not, however, the interface that solely governs the fragmentation pattern. Relevant simulations [1] suggest, in addition to the importance of the interface properties discussed in the next paragraph, that: (a) the fiber fragmentation pattern depends strongly on the volume of the matrix present, (b) the matrix hardening modulus E_{mh} influences the fragmentation pattern, (c) the overall load-displacement response of the specimen is practically insensitive to the interface properties, (d) it is practically impossible to achieve average fragmentation lengths of the order of or lower than the fiber diameter.

In the numerical analysis, the transition region between fiber and matrix is modeled as an "interphase." Two values are necessary to describe its elastic brittle-

response - the interphase stiffness coefficient S (modulus over thickness) and the failure stress coefficient F (failure stress over thickness). From back-analysis of the fragmentation test, the values for Ti-6Al-4V matrix, SiC (SCS-6) fiber composite were identified as: $S=860$ MPa/fm, $F=11$ MPa/fm. It is noted that the fragmentation pattern depends greatly on F , and not as much on S . For example, the fragmentation pattern obtained with the above values [1], is insensitive (within 100% changes) to the value of S , but depends strongly on F .

2. PUSHOUT AND PULLOUT TESTS

Such tests are commonly performed on metal as well as ceramic matrix composites. The specimens contain a single fiber with its ends exposed at both ends. In the pushout test an indenter loads the fiber in compression until it slides out of the matrix. The pullout test is similar, with the difference that tensile load is applied on the fiber. For metal matrix composites our simulations showed the following: (a) small, stable cracks form at the interface near the external load application side, for both the pushout and pullout tests, (b) after the arrest of the cracks mentioned above, tension and shear dominated interface cracks initiate at the side opposite to the external load application, (c) after the arrest of the cracks that initiated at the bottom, further external load is required for the whole interface to fail, (d) the size of the "hole" at the bottom support has significant effect on the load-deformation response, and (e) the effects of residual stresses seem to be important. Details can be found in the literature [1].

3. TRANSVERSE TESTS

The transverse test which is used to study the interface fracture behavior is usually performed using single fiber model composites. Single fiber specimens are well suited for interface evaluation because, (a) the interface chemical bonding, which depends on the chemical reaction between the matrix and the fiber materials during processing, remains the same in a single fiber sample as well as in a multi-ply composite panel, and (b) the residual stresses at the interface are relatively easy to calculate in the single fiber sample, thereby making it feasible to account for the residual stresses in the modeling of the test to derive more accurate conclusions about the stress at fracture of the fiber-matrix interface. In the following we first describe some experimental results from recent transverse tests [2] and then the results from the lattice modeling studies.

Monofilament composite samples for these experiments were processed by the foil-fiber-foil technique wherein two Ti-6Al-4V sheets with a single SiC fiber (SCS-6 or SCS-0) between them were hot pressed at a condition of 960 °C at 17 MPa for 1.5 hours. The interfacial microstructure obtained with this processing condition clearly indicates that some chemical reaction between the graded carbon coating and the matrix has taken place during the consolidation process. Furthermore, this reaction is nonuniform. After processing of the single fiber composite samples, the samples are cut into dog-bone-shaped specimens with the fiber axis perpendicular to the loading axis of the samples as shown in Figure 1a.

3.1 Experimental Approach

Transverse tensile tests were carried out using a micro-straining stage built in the WL/Materials Directorate [3]. The loading was applied stepwise so that the ultrasonic scanning could be done in-situ under the loaded condition at different stress levels. An in-situ ultrasonic nondestructive technique was used for this purpose [3].

The damage of the fiber-matrix interface during the transverse test was evaluated using the shear-wave back reflectivity technique (SBR) [4, 5]. The loading of the

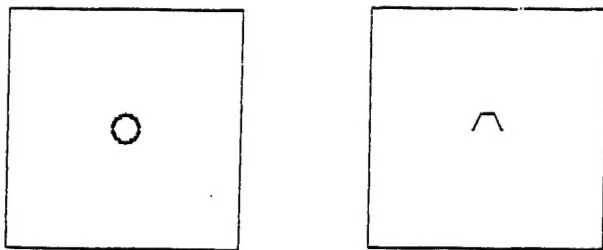
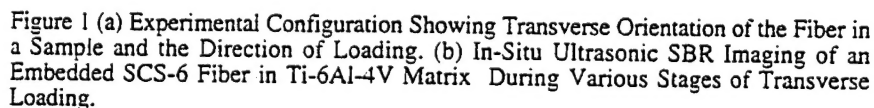
samples was done in incremental steps. At each step of loading, the fiber-matrix interface was imaged (while holding the sample under that load) in a pulse-echo mode using a focused ultrasonic beam which was incident on the surface of the samples at an angle of 24° , an angle between the first and the second critical angles of the matrix material. Hence, mode-converted vertically polarized shear waves were incident on the fiber-matrix interface. The shear waves were back-reflected to the transducer and the reflection coefficient from the interface was evaluated to characterize the interface fracture. The model presented in literature [4, 5] was used for the modeling of the reflection coefficient and the interface fracture was modeled based on theoretical predictions and experimental data. The reflected amplitude from a hole, which represents a complete debond (see previous section), was used to calibrate the maximum reflected signal from the fiber to correspond to the fracture of the interface.

Figure 1b shows the stress-strain diagram for the tensile test of a Ti-6Al-4V/SCS-6 single fiber composite sample under transverse loading, and the corresponding ultrasonic shear wave images at various stress levels of the sample labeled 'A' through 'K' in the Figure. The image labeled 'A' in Figure 1b corresponds to the fiber-matrix interface before the commencement of the mechanical loading of the sample. Reflectivity image 'B' indicates the first few points of interface that fracture at about 350 MPa (shown with the maximum amplitude calibrated to red in the color scale). These points of interface fracture are located near the two ends of the fiber and in several places at the center of the fiber. The reflectivity images 'C' through 'J' show the progression of interfacial damage as the load increases. The image 'K' shows that the entire interface has been fractured at about 700 MPa. An important conclusion drawn from using in-situ SBR imaging of the transverse test contradicts a commonly accepted assumption that the entire interface is likely to fracture almost instantaneously once a sufficiently high stress level is reached because of the existence of a weak diffusion bond [6] (as contrasted to mechanical bonding). The work reported here suggests that the debonding progresses from a small number of isolated points at a low load, to the entire interface over a finite range of applied stress (almost 350 MPa as shown in Figure 1b). This range is dependent mainly on the redistribution of stress along the interface which occurs due to the propagation of the interfacial crack as well as on the homogeneity of the fiber-matrix interface. Finally, the applied load completely released and the image of the fiber significantly reduced as the open interface cracks closed.

3.2 Lattice Modeling of Transverse Tests

The test configurations mentioned above are three-dimensional. Thus a simulation should reflect the 3-D effects. In [1] the simulations were two-dimensional; the limitations from such a "simplification" are discussed extensively in [1]. For the fragmentation and pullout, pushout tests, since the load is applied parallel to the fiber, the limitations from a 2-D simulations may not be of great importance, although a 3-D simulation process would be preferable. The reason for performing 2-D simulations have been the following: (a) there are several issues to be resolved, i.e. thermally induced residual stresses, the implications of assuming a homogenized interphase, effects of temperature on interface properties, imperfections in the interface, fiber, and matrix, etc., before a fully 3-D simulation process is employed, (b) for 3-dimensional analysis, only problems with rather coarse discretization can be solved within reasonable computer time use. For the transverse test, the limitations of a 2-dimensional analysis may be important. However, certain issues can be examined by analyzing a configuration like the one shown in Figure 2a.

Figure 2b shows the crack pattern obtained from the lattice analysis, at the load level shown by an arrow in Figure 3. The following have been identified: (a) before the matrix yields, the interface starts to fail, (b) interface failure is not symmetric with respect to either the loading axis, or the axis perpendicular to the loading direction, (c) although interface failure is locally brittle, the final stage of interface failure is reached



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In our simulations interface failure initiated at about 339 MPa external stress and progressed up to about 520 MPa. Thus it is difficult to define an external "stress" that corresponds to interface failure for this experimental configuration. Despite this difficulty in order to examine the response along the direction of the fiber, we have considered a "failure stress" of 420 MPa and with this value we performed simulations (2-D), as shown in Figure 5a, where now the x-axis is parallel to the fiber and the y-axis coincides with the external load axis. In these simulations, all interface elements are subjected to the same stress, thus we chose to initiate fracture at the right free-surface. We believe that residual stresses, which the present analysis does not consider, will be important for interface fracture along the fiber direction. Figure 5b shows the interface crack propagation at some point of external load application. Here interface fracture is "brittle" (the external stress increase from interface fracture initiation to its complete fracture is not appreciable).

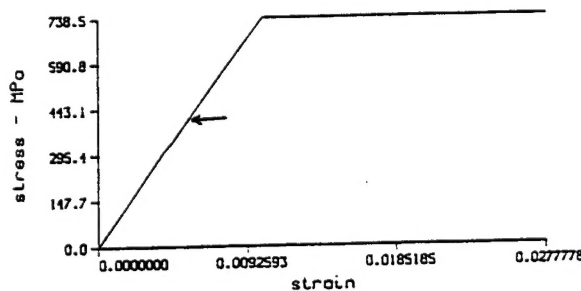


Figure 3: The external stress Vs external strain response as calculated from the two-dimensional simulation.

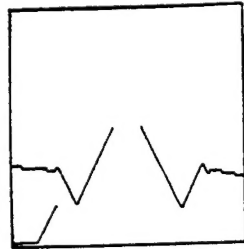


Figure 4: Simulation results. Failure (yielding) of the ductile matrix, occurring after interface failure. The corresponding external strain is 0.02.

Thus, for the transverse test, interface (overall) failure on the plane transverse to the fiber axis is non-brittle, while crack propagation in the direction of the fiber is rather brittle. This shows the strong interactions among the different directions. In other words, three-dimensional effects are important for the transverse test. Mean interface failure values can be obtained, however, by 2-D analysis using the configuration of Figure 2. It is once again noted that the simulations were performed using interface values as determined from the back analysis of the fragmentation test. Thus the results of the 2-D simulations can be compared with the mean (averaged over the fiber direction)

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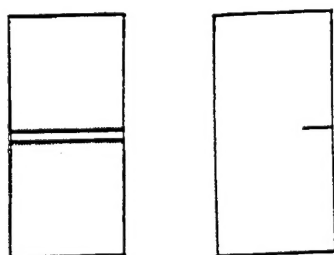


Figure 5: (a) Along the fiber axis two-dimensional configuration of the transverse test. The bars in the lattice assigned interface properties are shown. The fiber is inside the two parallel lines and the rest of the configuration is matrix. (b) Interface crack pattern soon after crack initiation.

ACKNOWLEDGMENTS

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